ATTACHMENT III –

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Aluminum and Aluminum Alloys

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rate estimation is that the only effect of temperature is on the kinetics of precipitation. This assumption is not valid, however, when portions of the metal are quenched locally but reheated significantly before quenching is complete.

Age Hardening

After solution treatment and quenching, hardening is achieved either at room temperature (natural aging) or with a precipitation heat treatment (artificial aging). In some alloys, sufficient precipitation occurs in a few days at room temperature to yield stable products with properties that are adequate for many applications. These alloys sometimes are precipitation heat-treated to provide increased strength and hardness in wrought or cast products. Other alloys with slow precipitation reactions at room temperature are always precipitation heat-treated before being used.

In some alloys, notably those of the 2xxx series, cold working of freshly quenched material greatly increases its response to later precipitation heat treatment. Mills take advantage of this phenomenon by applying a controlled amount of rolling (sheet and plate) or stretching (extrusion, bar, and plate) to produce higher mechanical properties. However, if the higher properties are used in design, reheat treatment must be avoided.

Natural Aging. The more highly alloyed members of the 6xxx wrought series, the copper-containing alloys of the 7xxx group, and all of the 2xxx alloys are almost always solution heat-treated and quenched. For some of these alloys, the precipitation hardening that results from natural aging alone produces useful tempers (T3 and T4 types) that are characterized by high ratios of tensile to yield strength, high fracture toughness, and high resistance to fatigue. For the alloys that are used in these tempers, the relatively high supersaturation of atoms and vacancies retained by rapid quenching causes rapid formation of GP zones, and strength increases rapidly, attaining nearly maximum stable values in four or five days. Tensile property specifications for products in T3- and T4-type tempers are based on a nominal natural aging time of four days. In alloys for which T3- or T4-type tempers are standard, the changes that occur on further natural aging are of relatively minor magnitude, and products of these combinations of alloy and tempers are regarded as essentially stable after about one week.

In contrast to the relatively stable condition reached in a few days by 2xxx alloys that are used in T3- or T4-type tempers, the 6xxx alloys

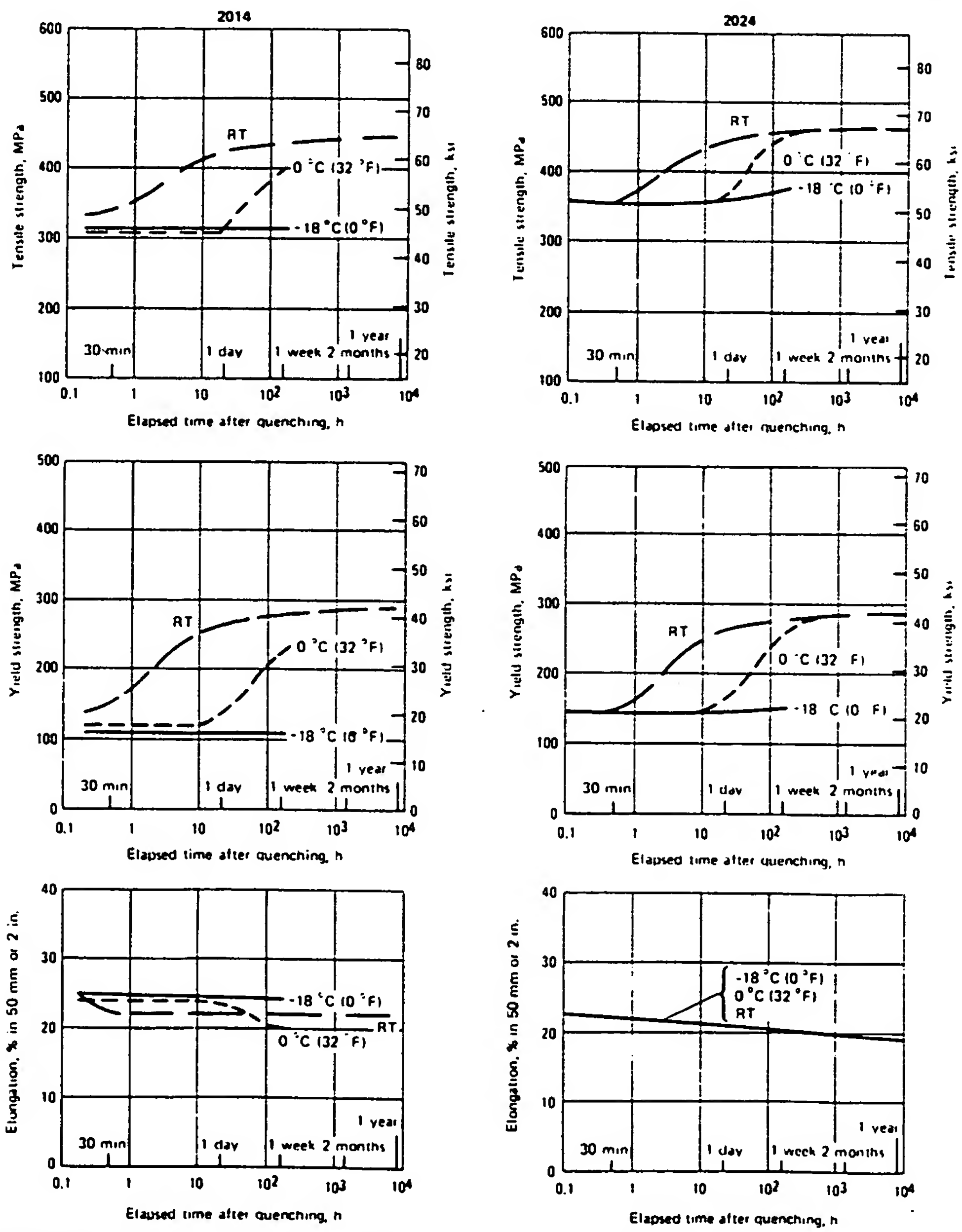

Fig. 22 Aging characteristics of aluminum sheet alloys at room temperature, at 0 °C (32 °F), and at -18 °C (0 °F). (continued)

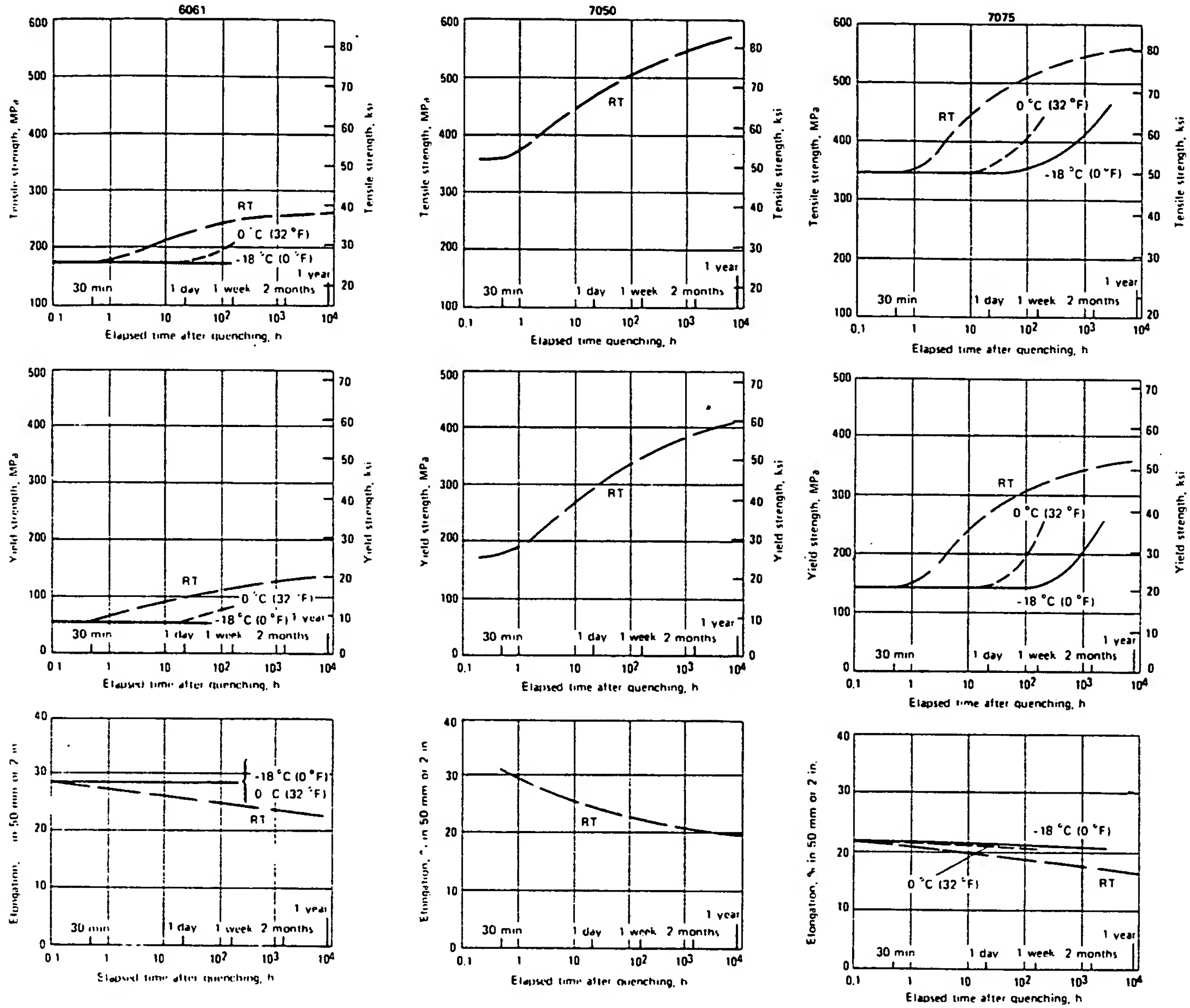

Fig. 22 (continued) Aging characteristics of aluminum sheet alloys at room temperature, at 0 °C (32 °F), and at −18 °C (0 °F)

and to an even greater degree the 7*xxx* alloys are considerably less stable at room temperature and continue to exhibit significant changes in mechanical properties for many years. The differences in rate and duration of changes in tensile yield strength of representative alloys of the three types are illustrated in Fig. 22. Because of the relative instability of the 7*xxx* alloys, the naturally aged temper (after solution heat treatment and quenching) is designated by the suffix letter W. For a specific description of this condition, the time of natural aging should be included (example: 7075-W, 1 month).

Aging characteristics vary from alloy to alloy with respect to both time to initial change in mechanical properties and rate of change, but aging effects always are lessened by reductions in aging temperature (see Fig. 22). With some alloys, aging can be suppressed or delayed for several days by holding at a temperature of −18 °C (0 °F) or lower. It is usual practice to complete forming and straightening before aging changes mechanical properties appreciably. When scheduling makes this impractical, aging may be avoided in some alloys by refrigerating prior to forming. It is conventional practice to refrigerate alloy 2024-T4 rivets to maintain good driving characteristics. Full-size wing plates for current-generation jet aircraft have been solution heat-treated and quenched at the primary fabricating mill, packed in dry ice in specially designed insulated shipping containers, and transported by rail about 2000 miles to the aircraft manufacturers plant for forming.

Unanticipated difficulties may arise as a result of failure to control refrigerator or part temperature closely enough. If opening of the cold box to insert or remove parts is done too frequently, the cooling capacity of the refrigerator may be exceeded. At times, the rate at which heavy-gage parts can be cooled in a still-air cold box has been found to be insufficient. This problem has been solved in one plant by immersing parts in a solvent at −40 °C (−40 °F) before placing them in the refrigerator.

The T3-type tempers are distinguished from T4-type tempers by significant mechanical property differences that result from cold work strain hardening associated with certain mechanical operations performed after quenching. Roller or stretcher leveling to achieve flatness or straightness introduces modest strains (of the order of 1 to 4%) that cause changes in mechanical properties (primarily, increases in strength). Further increases in strength can be obtained by cold rolling, additional stretching, combinations of these operations, or (for products such as hand forgings) compressive deformation. These operations followed by natural aging alone (no precipi-

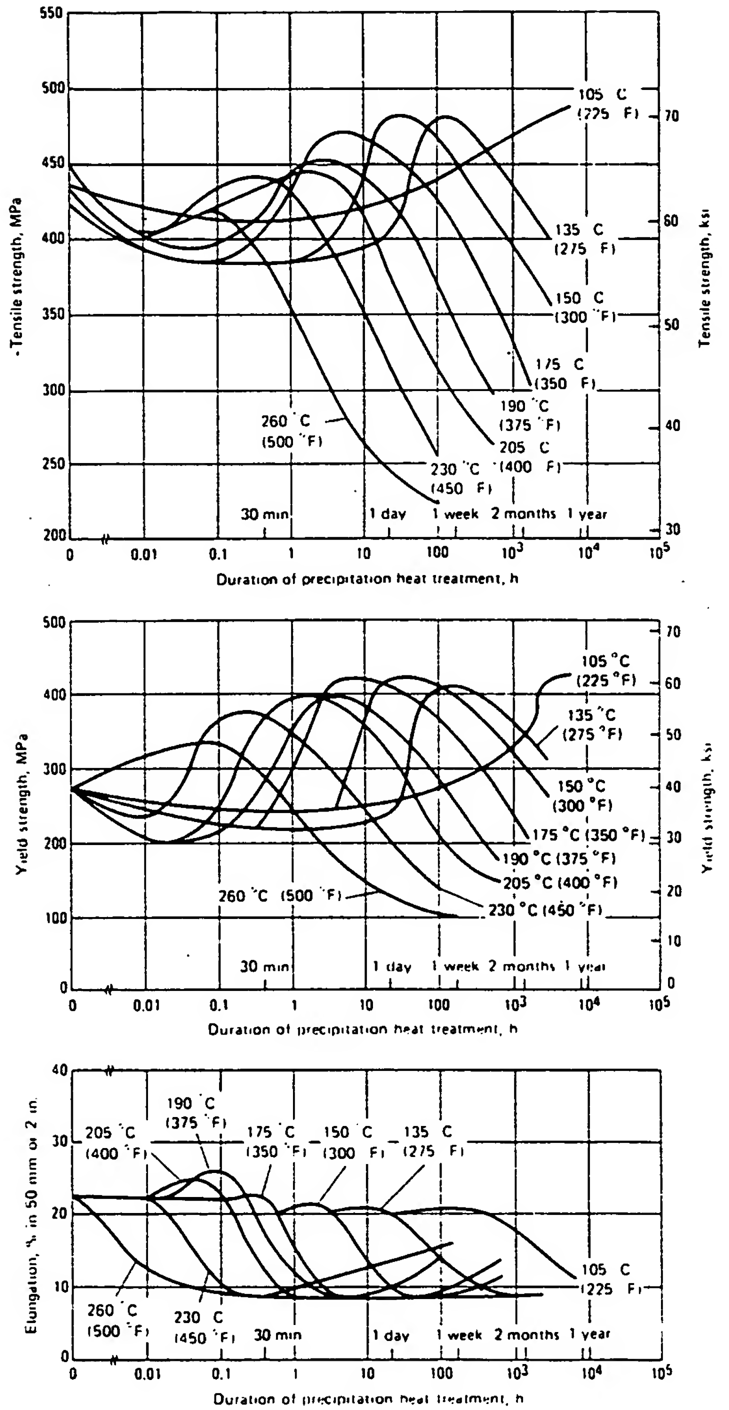

Fig. 23(a) Aging characteristics of alloy 2014 sheet

tation heat treatment) produce T3-type tempers. An additional digit is used to indicate a variation in strain hardening that results in significant changes in properties. In the most recently introduced 2xxx aircraft alloy, 2324, high strength is achieved by cold-rolling plate to a T39 temper.

Precipitation heat treatments generally are low-temperature, long-term processes. Temperatures range from 115 to 190 °C (240 to 375 °F); times vary from 5 to 48 h.

Choice of time-temperature cycles for precipitation heat treatment should receive careful consideration. Larger particles of precipitate result from longer times and higher temperatures; however, the larger particles must, of necessity, be fewer in number with greater distances between them. The objective is to select the cycle that produces the optimum precipitate size and distribution pattern. Unfortunately, the cycle required to maximize one property, such as tensile strength, is usually different from that required to maximize others, such as yield strength and corrosion resistance. Consequently, the cycles used represent compromises that provide the best combinations of properties.

Production of material in T5- through T10-type tempers (see the article "Alloy and Temper Designation Systems" in this Volume) necessitates precipitation heat treating at elevated temperatures (artificial aging). Although the hardening precipitate developed by this operation is submicroscopic, structures before and after precipitation heat treatment often can be distinguished by etching metallographic specimens. In aluminum alloys in the solution heat-treated and quenched condition, the color contrast between grains of differing orientations is relatively high, particularly in 2xxx series wrought alloys and 2xx.0 series casting alloys. The contrast is noticeably decreased by precipitation heat treatment.

Differences in type, volume fraction, size, and distribution of the precipitated particles govern properties as well as the changes observed with time and temperature, and these are all affected by the initial state of the structure. The initial structure may vary in wrought products from unrecrystallized to recrystallized and may exhibit only modest strain from quenching or additional strain from cold working after solution heat treatment. These conditions, as well as the time and temperature of precipitation heat treatment, affect the final structure and the resulting mechanical properties.

Because the mechanical properties and other characteristics change continuously with time and temperature, as shown in Fig. 23(a) to 23(d) for three wrought alloys, treatment to produce a combination of properties corresponding to a specific alloy-temper combination requires one or more rather specific and coordinated combinations of time and temperature. Both parameters are subject to practical limitations. Recommended commercial treatments often are compromises between time/cost factors and the probability of obtaining the intended properties, with allowances for variables such as composition within specified ranges and temperature variations within the furnace and load. Use of higher temperatures may reduce treatment time, but if the temperature is too high, characteristic features of the precipitation hardening process reduce the probability of obtaining the required properties.

T6 and T7 Tempers. Precipitation heat treatment following solution heat treatment and quenching produces T6- and T7-type tempers. Alloys in T6-type tempers generally have the highest strengths practical without sacrificing the minimum levels of other properties and characteristics found by experience to be satisfactory and useful for engineering applications. Alloys in T7-type tempers are overaged, which means that some degree of strength has been sacrificed to improve one or more other characteristics. Strength may be sacrificed to improve dimensional stability, particularly in products intended for service at elevated temperatures, or to lower residual stresses in order to reduce warpage or distortion in machining. T7-type tempers fre-

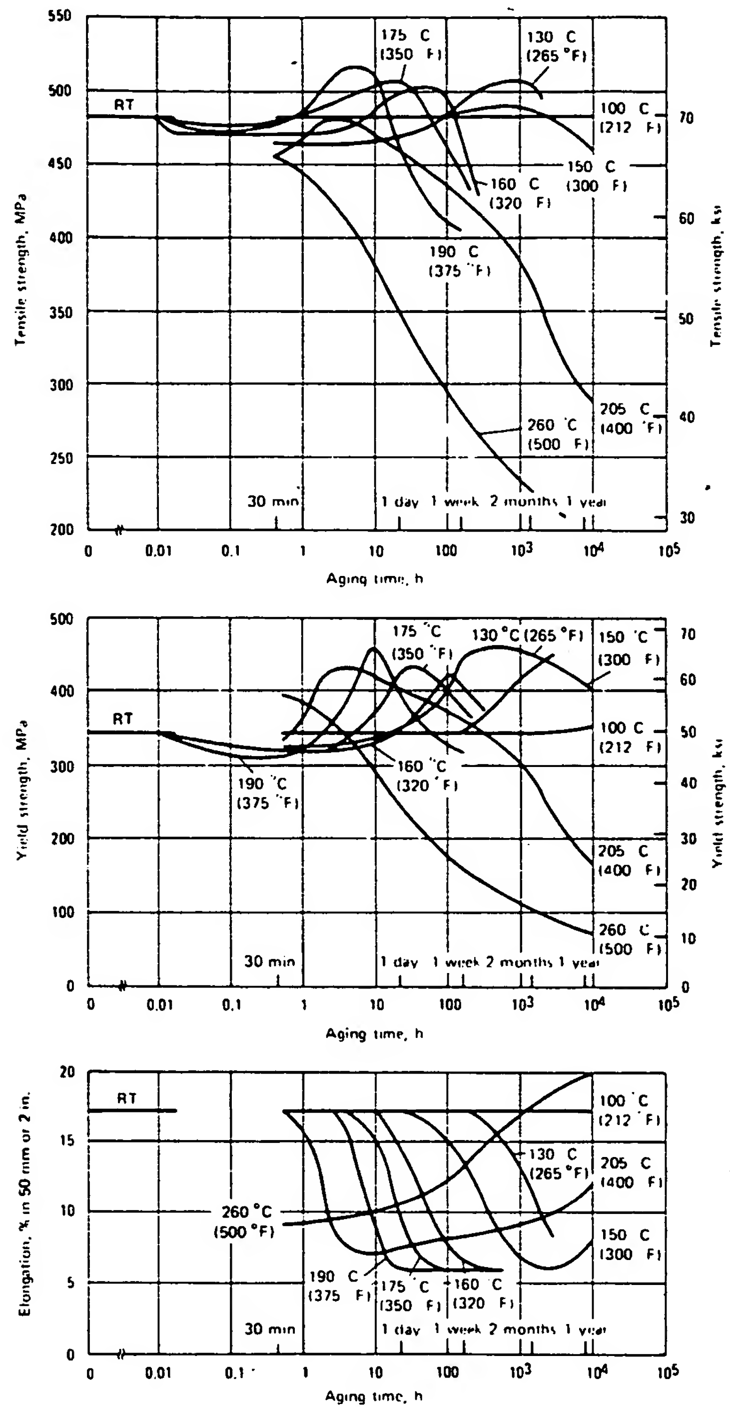

Fig. 23(b) Aging characteristics of alloy 2024 sheet (see also Fig. 23d)

quently are specified for cast or forged engine parts. Precipitation heat treating temperatures used to produce these tempers generally are higher than those used to produce T6-type tempers in the same alloys.

Two important groups of T7-type tempers, T73 and T76, have been developed for the wrought alloys of the 7*xxx* series, which contain more than about 1.25% copper. These tempers are intended to improve resistance to exfoliation corrosion and stress-corrosion cracking, but as a result of overaging, they also increase fracture toughness and, under some conditions, reduce rates of fatigue crack propagation. The T73 temper has greatly minimized stress-corrosion cracking of large and complex machined parts made of these alloys, which occasionally occurred with T6-type tempers. Additional information on the effect of various tempers on the corrosion resistance of aluminum alloys can be found in the article "Corrosion Behavior" in this Volume.

The precipitation heat treatment used to produce the T73 and T76 tempers consists either of a two-stage isothermal precipitation heat treatment or of heating at a controlled rate to a single treatment temperature. The microstructural/electrochemical relationships that are required in order to achieve the desired corrosion-resisting characteristics can be developed by using only a single-stage precipitation heat treatment above about 150 °C (300 °F), but higher strength is obtained by preceding this with a lower-temperature stage or with a slow-controlled heatup. Extended natural aging can provide the same results, but the times required at room temperature are impractical. Either during the preliminary stage or during slow heatup, a fine, high-density dispersion of GP zones is nucleated. Either the time and temperature of the first step or the rate of heating must be controlled to produce GP zones that will not dissolve but will transform to the ′ precipitate when heated to the, aging temperature above 150 °C (300 °F). The aging practice that produces the results in the shortest time depends on the GP-zone solvus temperature. This temperature, in turn, depends on vacancy concentration (a factor influenced by solution heat-treating temperature and quench rate) and on composition. If first-step aging time is too short, if first-step aging temperature is too far below the GP-zone solvus, or if heating rates are too high, the GP zones will dissolve above 150 °C (300 °F), and the resultant coarse and widely distributed precipitate will provide lower strength. The T76-temper treatments have the same operational sequence but employ second-stage heating only long enough to develop a resistance to exfoliation corrosion higher than that provided by the T6-type tempers. Materials in the T73 temper also have high resistance to exfoliation corrosion.

Recommended treatments to produce T5- and T6-type tempers, and those of the T7-type employed for dimensional and property stabilization, provide adequate tolerance for normal variations encountered with good operating practices. On the other hand, the T73, T74 (formerly T736), and T76 tempers for alloys 7049, 7050, 7075, 7175, and 7475 involve changes in strength that occur significantly more rapidly at the temperatures employed in the second stage of the T7*x* precipitation heat treatment cycle compared to the changes occurring at the temperatures employed to produce the T6 temper.

As illustrated in Fig. 24, variations in soak time of several hours, and variations in soak temperature of up to 11 °C (20 °F) from the nominal aging practice of 24 h at 120 °C (250 °F), affect the strength of 7075-T6 by as much as 28 MPa (4 ksi). In contrast, similar variations in second-step soak time and temperature for 7075-T73—that is, variations for 24 h at 165 °C (325 °F)—affect strength by up to 150 MPa (22 ksi).

Consequently, control of both temperature and time to achieve the mechanical properties and corrosion resistance specified for these tempers is more critical than the control required in producing the T6-type temper. More-

over, rate of heating from the first to the second aging step must be considered, because precipitation occurs during this period.

Heat treaters attempt to overcome these problems by empirically modifying soak times to compensate for precipitation during heating and for the effects of soaking at temperatures above or below the nominal. A method has been developed (Ref 8) that permits quantitative compensation for the effects of precipitation during heating and the effects of soaking either above or below the recommended temperature. For overaging, these effects can be described by the following equation:

$$YS = Y \exp - \left(\frac{t_c}{F_{YS}} + \theta \right) \quad \text{(Eq 5)}$$

where YS is yield strength; Y is a term having units of strength that is dependent on alloy, fabrication, and test direction; t_c is time at soak temperature; F_{YS} is a temperature-dependent term; and

$$\theta = \int \frac{dt}{F_{YS}} \quad \text{(Eq 6)}$$

where t is time during heating.

Equation 5 provides the basis for selecting a nominal aging time that will result in the desired yield strength. It gives the furnace operator a method of compensating for heating rate and for differences between desired and attained soak temperatures.

Specifics will be illustrated using data for alloy 7050. The value of F_{YS} (in hours) for 7050 can be calculated by the following equation:

$$F_{YS} = 1.45 \times 1^{-16} \exp \left(\frac{32{,}562}{T_F + 460} \right) \quad \text{(Eq 7a)}$$

where T_F is temperature in °F, or

$$F_{YS} = 1.45 \times 10^{-16} \exp \left(\frac{18{,}090}{T_K} \right) \quad \text{(Eq 7b)}$$

where T_K is temperature in K.

In one experiment, lengths of 7050-W (4 days) extrusions were aged at 24 h at 120 °C (250 °F) plus the equivalent of 3 to 42 h at 165 °C (325 °F). For the second step, a logarithmic heatup was used in which 10 h were required for the load to reach 155 °C (315 °F) and the nominal soak temperature was 165 °C (325 °F). Figure 25 indicates that yield strength generally agreed with the values predicted using Eq 5. The deviation of the curve for short-transverse strength at the short aging times indicates that the method is inadequate for predicting strength on the underaging side of the aging curve.

The effects of neglecting to compensate for soaking at temperatures other than the nominal can be large (Fig. 26). For example, the calculated difference in strength between alloy 7050 extrusions soaked 29 h at 160 °C (320 °F) and at 165 °C (325 °F) is about 50 MPa (7 ksi), and the calculated difference in strength between 7050 extrusions soaked 29 h at 155 °C (315 °F) and at 170 °C (335 °F) is about 100 MPa (14 ksi).

Neglecting to compensate for time spent heating the work to the soak temperature will increase the variability. Strength loss attributed to heatup was 14 MPa (2 ksi).

These kinetic relationships also can assist in selection of equivalent aging times for alter-

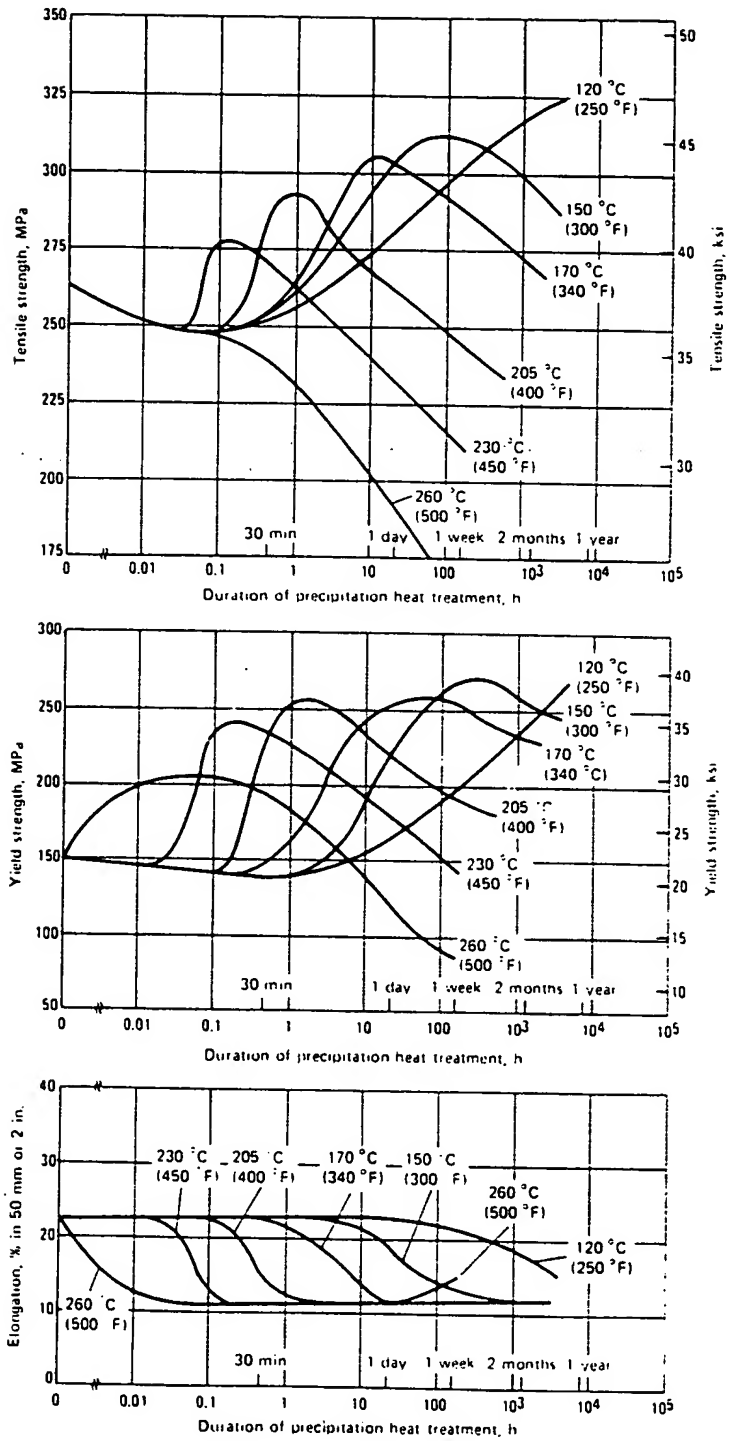

Fig. 23(c) Aging characteristics of alloy 6061 sheet

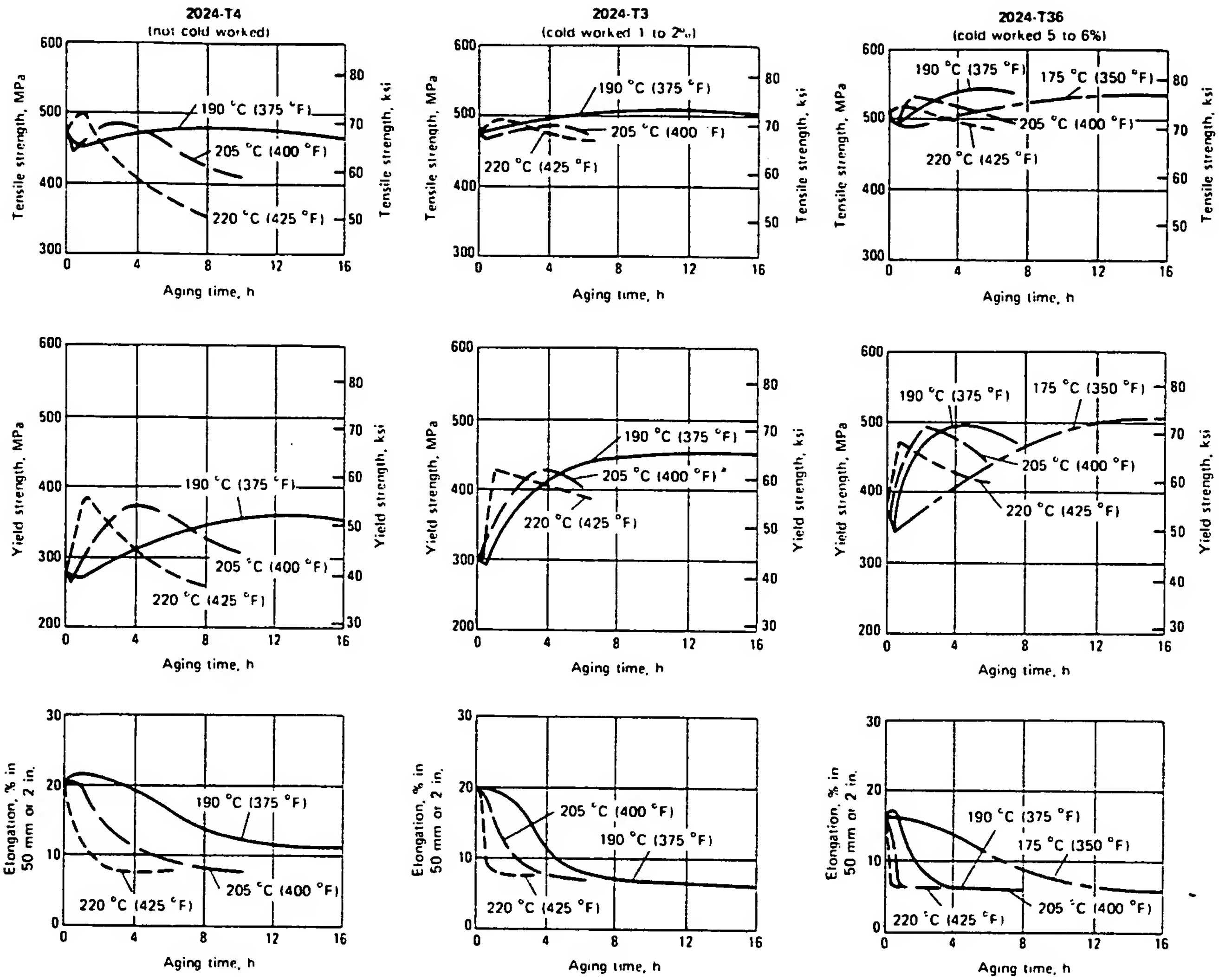

Fig. 23(d) Effects of cold work after quenching and before aging on tensile properties of alloy 2024 sheet

nate second-step aging temperatures. Equations 5 and 7 can be rearranged to yield the following equation:

$$t_2 = t_1 \exp \frac{32{,}562}{T_1 + 460} - \frac{32{,}562}{T_2 + 460} \quad \text{(Eq 8)}$$

where t_1 is aging time at temperature T_1, t_2 is aging time at temperature T_2 that will provide equivalent yield strength, and T_1 and T_2 are in °F. For example, the time at 175 °C (350 °F) equivalent to aging alloy 7050 for 29 h at 165 °C (325 °F) is calculated as follows:

$$t_{350} = 29 / \exp(1.28) = 29 / 3.6 = 8 \text{ h}$$

Thermomechanical effects on aging occur from deformation after solution heat treatment. The deformation step may be warm or cold, and it may occur before, after, or during aging. The simplest thermomechanical practices are those of the conventional T3-type, T8-type, or T9-type tempers. The rate and extent of precipitation strengthening are distinctly increased in some alloys by cold working after quenching, whereas other alloys show little or no added strengthening when treated by this sequence of operations.

Alloys of the 2*xxx* series, such as 2014, 2124, and 2219, are particularly responsive to cold work between quenching and aging, and this characteristic is the basis for the higher-strength T8-type tempers. The strength improvement accruing from the combination of cold working and precipitation heat treating is a result of nucleation of additional precipitate particles by the increased strain. In some alloys of the 2*xxx* series, strain introduced by cold working after solution heat treatment and quenching also incudes nucleation of a finer precipitate dispersion that increases strength. Depending on the temper, however, toughness may be adversely affected, as illustrated in Fig. 27 for 2024 sheet.

Strengthening from thermomechanical processing is the basis for the higher-strength T8-type tempers of alloys 2011, 2024, 2124, 2219, and 2419, which are produced by applying controlled amounts of cold rolling, stretching, or combinations of these operations. Normally, cold work is introduced by stretching; however, other methods can be used, such as cold rolling. Recently, 2324-T39 was developed. The T39 temper is obtained by cold rolling approximately 10% after quenching, followed by stretching for stress relief. This type of approach results in strengths similar to those obtained with T8 processing, but with the better toughness and fatigue characteristics of T3-type products. Alloys 2024, 2124, and 2219 in T8-type tempers are particularly well suited for supersonic and military aircraft; alloy 2219 in such tempers and alloy 2014-T65 were the principal materials for the fuel and oxidizer tanks (which also served as the primary structure) of the Saturn V space vehicles. Re-solution heat treatment of mill products supplied in these tempers can result in grain growth and in substantially lower strength than is normal for the original temper. Such reheat treatment is not recommended.

Alloys of the 7*xxx* series do not respond favorably to the sequence of operations used to

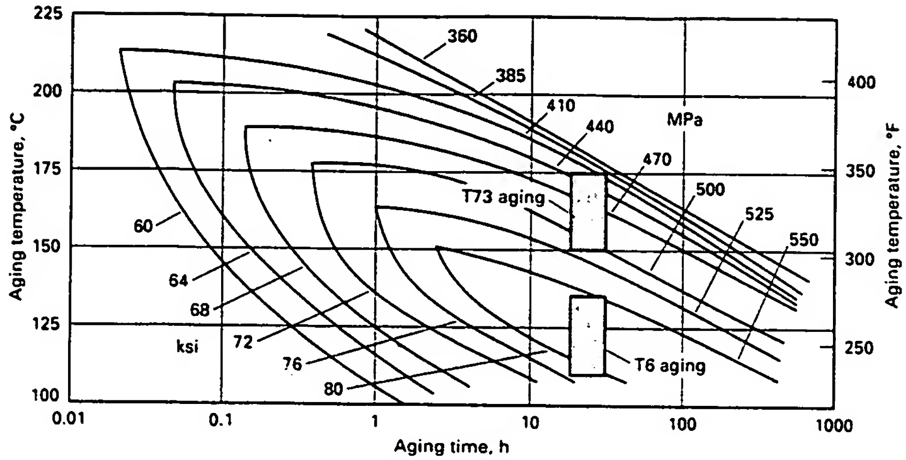

Fig. 24 Iso-yield-strength curves for alloy 7075

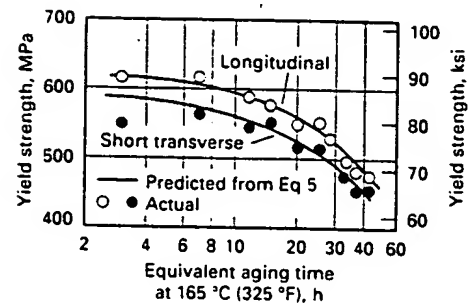

Fig. 25 Actual versus predicted yield strengths for alloy 7050 extrusions

produce T8-type tempers, and no such tempers are standard for these alloys. The strains associated with stretching or compressing of 7xxx alloys have relatively little effect on the mechanical properties of material that is precipitation heat-treated to T6-type tempers. On the other hand, these operations have measurable detrimental effects on final strength when T73, T736, or T76 tempers are produced, particularly in the direction opposite the direction of cold work. Accordingly, specification properties are somewhat lower for the stress-relieved versions of these tempers. Decreasing the overaging time to compensate for the loss in strength is not advisable, because this would impair development of the desired corrosion characteristics.

Temperature control and uniformity present essentially the same problems in precipitation heat treating as they do in solution heat treating. Good temperature control and uniformity throughout the furnace and load are required for all precipitation heat treating. Recommended temperatures are generally those that are least critical and can be used with practical time cycles. Except for 7xxx alloys in T7-type tempers, these temperatures generally allow some latitude and should have a high probability of meeting property specification requirements. Furnace radiation effects seldom are troublesome except in those few furnaces that are used for both solution and precipitation heat treating. Generally, such situations should be avoided, because the high heat capacity needed for the higher temperatures may be difficult to control at normal aging temperatures.

Soak time in precipitation heat treating is not difficult to control; the specified times carry rather broad tolerances. Heavier loads with parts racked closer together, and even nested, are not abnormal. The principal hazard is undersoaking due to gross excesses in loading practices. Some regions of the load may reach soak temperature long after soak time has been called. Placement of load thermocouples is critical, and limiting the size and spacing of a load may be necessary for aging to the T73 and T76 tempers. As discussed above, soak time is not as critical for peak-aged (T6- and T8-type) tempers.

Hardening of Cast Alloys

In general, the principles and procedures for heat treating wrought and cast alloys are similar. The major differences between solution treating conditions for castings and those for wrought products are found in soak times and quenching media. Solution of the relatively large microconstituents present in castings requires longer soaking periods than those used for wrought products (Table 5). When heat treatment of castings must be repeated, solution times become similar to those for wrought products, because the gross solution and homogenization has been accomplished and is irreversible under normal conditions. Reduction of stresses and distortion from quenching is also important, because castings generally are complex shapes with variations in section thickness.

Different casting processes and foundry practices also result in microstructural differences with relevance to heat treatment practice, because the coarser microstructures associated with slow solidification rates require a longer solution heat treatment exposure. Therefore, the time required at temperature to achieve solution is progressively shorter for investment, sand, and permanent-mold castings. Foundry practice (chilled, gating, type of mold) also plays an important role in the response of a casting, or a portion of a casting, to heat treatment. For example, thin-wall sand castings produced with extensive use of chills can often display finer microstructures than heavy-section permanent-mold parts produced in such a way that process advantages are not exploited.

For these reasons, solution heat treatment practices can be optimized for any specific part to achieve solution with the shortest reasonable cycle once production practice is finalized, even though most foundries and heat treaters will standardize a practice with a large margin of safety. There also exists a fundamental difference between unmodified and modified alloys in which the size and shape of silicon crystals are modified with additions of elements such as calcium, sodium, strontium, or antimony. Modified alloys undergo rapid spheroidization, while complete spheroidization is not achieved in unmodified alloys, even after very long times. The practical implication is that shorter solution heat treatment could be employed in fully modified castings. The mi-

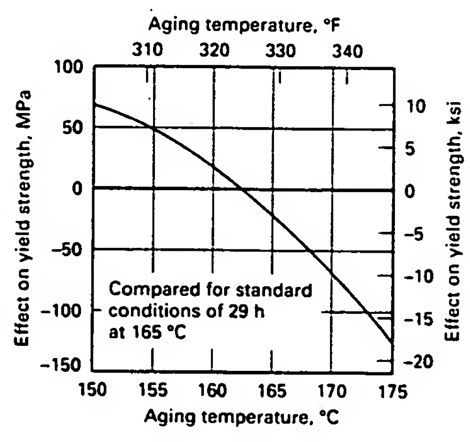

Fig. 26 Effect of aging temperature on yield strength of alloy 7050-T736

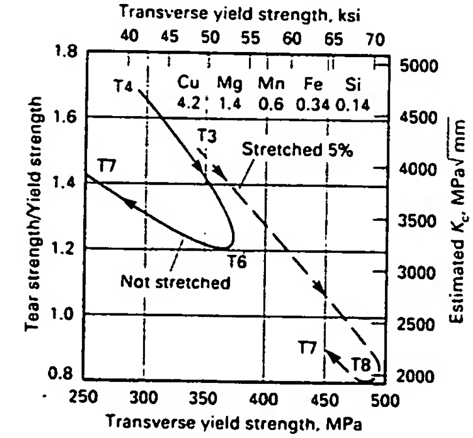

Fig. 27 Effect of stretching and aging on the toughness and yield strength of 2024 sheet

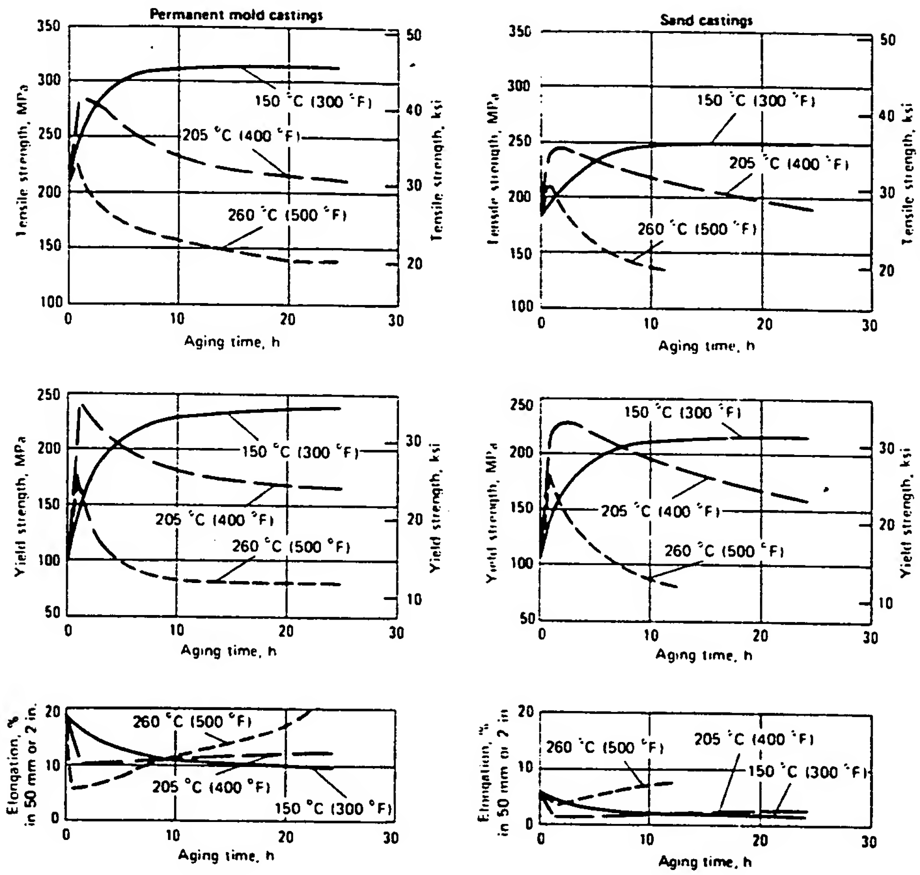

Fig. 28 Comparison of the precipitation-hardening characteristics of 356.0-T4 sand and permanent-mold castings

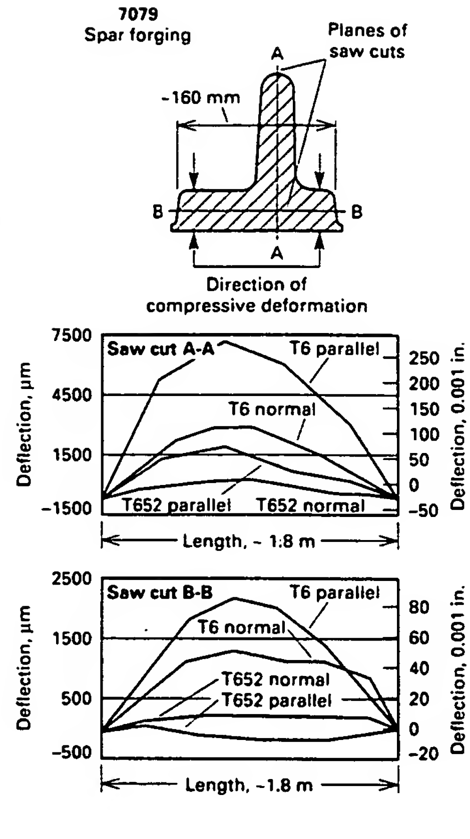

Fig. 29 Effect of 3% permanent deformation in compression (T652 treatment) on distribution of stress in a large forging. Parallel and normal refer to warpage directions with respect to the plane of the saw cut.

crosegregation of silicon and magnesium is not severe in the aluminum-silicon-magnesium casting alloys, and hence it takes only a short time to homogenize the alloy and to place the Mg_2Si into solution.

Quenchants. Quenching of aluminum castings is often done in boiling water or a milder medium to reduce quenching stresses in complex shapes. A commercially important variety is a mixture of polyalkylene glycol and water, which has no detrimental effect on properties for thicknesses under approximately 3.2 mm (0.125 in.). Quenchant additions can be made for the following purposes:

- To promote stable vapor film boiling by the deposition of compounds on the surface of parts as they are submerged in the quench solution
- To suppress variations in heat flux by increasing vapor film boiling stability through chemically decreased quench solution surface tension
- To moderate quench rate for a given water temperature

The key to the compromise between goals involving property development and the physical consequences of quenching is uniformity of heat extraction, which is in turn a complex function of the operable heat extraction mechanism. Nucleate, vapor film, and convective boiling occur with dramatically different heat extraction rates at different intervals. Differences in section thickness, load density, positioning, racking methods, surface condition, and casting geometry also influence the results.

Property Development. Yield strength is largely controlled by the limiting hardening-element level, and tensile strength (in a general sense) is related to the ductility at a given yield strength. Ductility, however, is controlled for a given yield strength by soundness and microstructural fineness, and it is thus determined in the foundry and not by the heat treater. This effect of casting methods on property development is shown in Fig. 28. Because of the finer cast structure and higher supersaturation of the more rapidly solidified permanent-mold castings, their tensile properties are superior to those of sand castings of the same composition similarly heat-treated.

Tempers. Cast products of heat-treatable aluminum alloys have the highest combinations of strength, ductility, and toughness when produced in T6-type tempers. Developing T6-type tempers in cast products requires the same sequence of operations employed in developing tempers of the same type in wrought products: solution heat treating, quenching, and precipitation heat treating. Premium-quality casting specifications such as MIL-A-21180 can require different strengths and ductility levels in the same casting.

Among precipitation treatments unique to castings are those resulting in the T5- and T7-type tempers. The T5-type temper is produced merely by applying a precipitation treatment to the as-cast casting, without previous solution treatment. A moderate increase in strength is achieved without warpage and subsequent straightening. High hardness, and dimensional and strength stability at elevated temperatures, account for the almost universal use of materials in T5-type tempers for pistons and other engine parts. Some applications demand combinations of strength, toughness, and dimensional stability that cannot be met by heat treating to T5, T6-, or T8-type tempers. For these applications, T7-type tempers are developed by solution heat treating, quenching in a medium that provides a moderate cooling rate, and then precipitation heat treating at a temperature higher than those used to develop T5-, T6-, and T8-type tempers. Heat treating to T7-type tempers results in lower strength than that of material in T6- or T8-type tempers, develops high ductility and

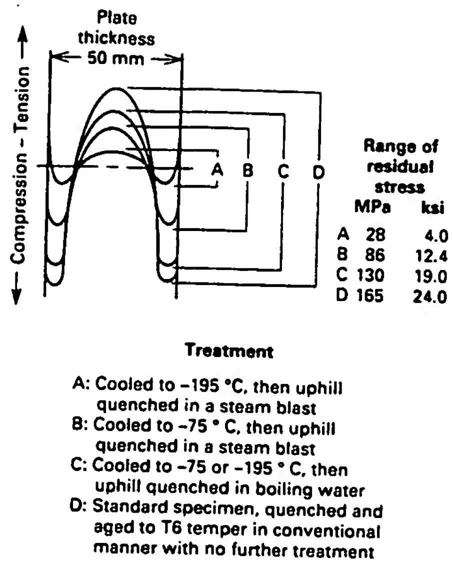

Fig. 30 Effectiveness of various uphill quenching treatments in reducing residual quenching stresses in 2014 plate. Uphill quenching treatments (single-cycle only) were applied from ½ to 1½ to 1 h after quenching from the recommended solution-treating temperature. All specimens were aged to the T6 temper after uphill quenching.

toughness, and carries precipitation far enough to minimize further precipitation during service.

Stress Relief

Immediately after being quenched, most aluminum alloys are nearly as ductile as they are in the annealed condition. Consequently, it is often advantageous to stress relieve parts by working the metal immediately after quenching. Numerous attempts also have been made to develop a thermal treatment that will remove, or appreciably reduce, quenching stresses. Normal precipitation heat-treating temperatures are generally too low to provide appreciable stress relief. Exposure to higher temperatures (at which stresses are relieved more effectively) results in lower properties. However, such treatments are sometimes used when even moderate reduction of residual stress levels is important enough so that some sacrifice in mechanical properties can be accepted. The T7-type temper for castings is a typical example of this kind of treatment.

Mechanical Stress Relief. Deformation consists of stretching (bar, extrusions, and plate) or compressing (forgings) the product sufficiently to achieve a small but controlled amount (1 to 3%) of plastic deformation. If the benefits of mechanical stress relieving are needed, the user should refrain from reheat treating.

Figure 29 illustrates the beneficial effect of 3% permanent deformation in compression on a large forging.

These methods are most readily adaptable to mill and forge shop products, as they require equipment of greater capacity than that found in most manufacturing plants. Application of these methods to die forgings and extrusions usually requires construction of special dies and jaws. Stretching generally is limited to material of uniform cross section; however, it has been applied successfully to stepped extrusions and to a 3 by 14 m (10 by 47 ft) aircraft wing skin roll-tapered to a thickness range of 7.1 to 3.2 mm (0.280 to 0.125 in.).

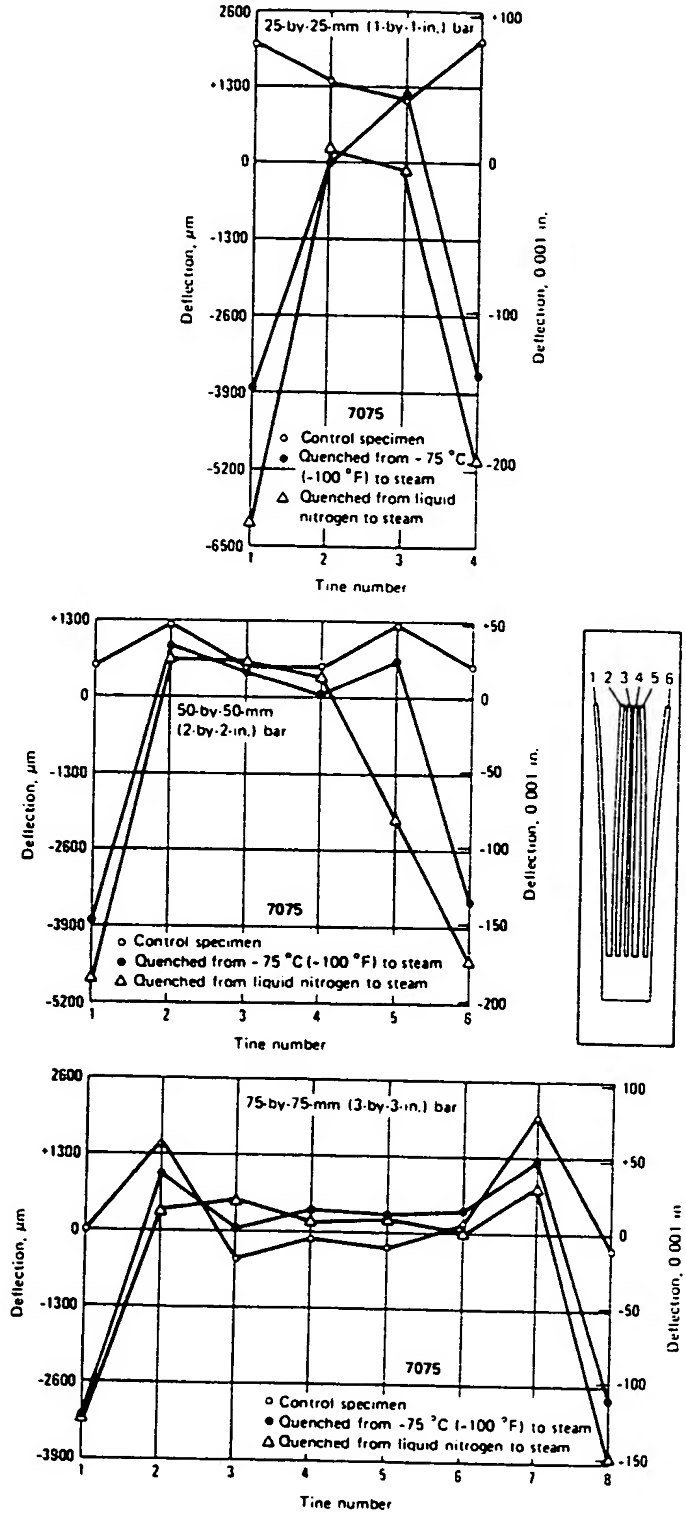

Fig. 31 Effect of uphill quenching on deflection of tines. Six-tine specimen was machined from 50 by 50 mm (2 by 2 in.) bar. Similar specimens machined from 25 by 25 mm (1 by 1 in.) and 75 by 75 mm (3 by 3 in.) bars had four and eight tines, respectively.

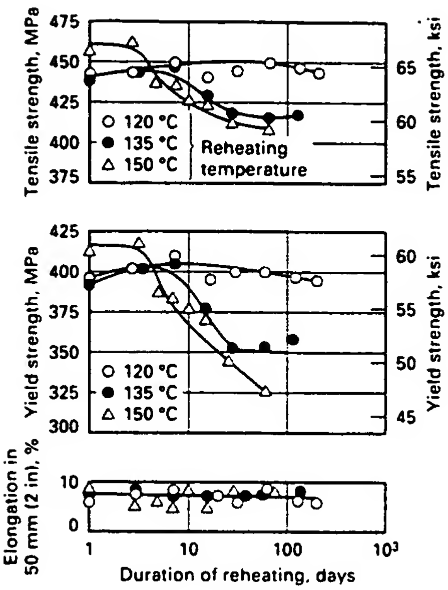

Fig. 32 Effects of reheating on tensile properties of Alclad 2024-T81 sheet

Table 11 Reheating schedules for wrought aluminum alloys
The schedules given in this table normally will not decrease strength more than 5%

Alloy and temper	Reheating time at a temperature of: 150 °C (300 °F)	165 °C (325 °F)	175 °C (350 °F)	190 °C (375 °F)	205 °C (400 °F)	220 °C (425 °F)	230 °C (450 °F)
2014-T4	(a)	(a)	(a)	(a)	(a)	(a)	(a)
2014-T6	2–50 h	8–10 h	2–4 h	½–1 h	5–15 min	(b)	(b)
2024-T3, 2024-T4	(a)	(a)	(a)	(a)	(a)	(a)	(a)
2024-T81, 2024-T86	20–40 h	...	2–4 h	1 h	½ h	15 min	5 min
6061-T6, 6062-T6, 6063-T6	100–200 h	50–100 h	8–10 h	1–2 h	½ h	15 min	5 min
7075-T6, 7178-T6	10–12 h	1–2 h	1–2 h	½–1 h	5–10 min	(b)	(a)

(a) Reheating not recommended. (b) Bring to temperature

Table 12 Effects of annealing treatments on ductility of 7075-O sheet

Annealing treatment	Elongation in tension(a), % in 50 mm (2 in.) for thickness of: 0.5 mm (0.020 in.)	1.6 mm (0.064 in.)	2.6 mm (0.102 in.)	Bend angle(b), degrees, for thickness of: 1.6 mm (0.064 in.)	2.6 mm (0.102 in.)	Elongation in bending(c), % in 50 mm (2 in.) for thickness of: 1.6 mm (0.064 in.)	2.6 mm (0.102 in.)
Treatment 1(d)	12	12	12	82	73	48	50
Treatment 2(e)	14	14	14	91	76	58	57
Treatment 3(f)	16	16	...	92.5	84	56	60

(a) Uniform elongation of gridded tension specimens. (b) Bend angle at first fracture. (c) Elongation in bend test for 1.3 mm (0.05 in.) gage spanning fracture. (d) Soak 2 h at 415 ± 14 °C (775 ± 25 °F); furnace cool to 260 °C (500 °F) at 30 °C/h (50 °F/h); air cool. (e) Soak 2 h at 425 °C (800 °F), air cool; soak 2 h at 230 °C (450 °F), air cool. (f) Soak 1 h at 425 °C (800 °F); furnace cool to 230 °C (450 °F) at 30 °C/h (50 °F/h); soak 6 h at 230 °C (450 °F), air cool

Specific combinations of the supplemental digits are used to denote the tempers produced when mechanical deformation is used primarily to relieve residual stresses induced during the quenching operation. For products stress-relieved by stretching, the digits 51 follow the basic T*x* designation (T451, for example). For products stress-relieved by compressive deformation, the supplementary digits are 52.

An additional digit is added to designations for extrusions: zero specifies that the product has not been straightened after final stretching; one indicates that straightening may have been performed after final stretching.

Effect of Precipitation Heat Treating on Residual Stress. The stresses developed during quenching from solution heat treatment are reduced during subsequent precipitation heat treatment. The degree of relaxation of stresses is highly dependent upon the time and temperature of the precipitation treatment and the alloy composition. In general, the precipitation treatments used to obtain the T6-type tempers provide only modest reduction in stresses, ranging from about 10 to 35%. To achieve a substantial lowering of quenching stresses by thermal stress relaxation, higher-temperature treatments of the T7 type are required. These treatments are used when the lower strengths resulting from overaging are acceptable.

Other thermal stress-relief treatments, known as subzero treatment and cold stabilization, involve cycling of parts above and below room temperature. The temperatures chosen are those that can be readily obtained with boiling water and mixtures of dry ice and alcohol—namely, 100 and −73 °C (212 and −100 °F)—and the number of cycles ranges from one to five. The maximum reduction in residual stress that can be effected by these techniques is about 25%. The maximum effect can be obtained only if the subzero step is performed first, and immediately after quenching from the solution treating temperature while yield strength is low. No benefit is gained from more than one cycle.

A 25% reduction in residual stress is sometimes sufficient to permit fabrication of a part that could not be made without this reduction. However, if a general reduction is needed, as much as 83% relief of residual stress is possible by increasing the severity of the uphill quench (that is, more closely approximating the reverse of the cooling-rate differential during the original quench). This may be accomplished by a patented process that involves extending the subzero step to −195 °C (−320 °F) and then very rapidly uphill quenching in a blast of live steam (Fig. 30). The rate of reheating is extremely critical; therefore, to ensure proper application of the steam blast, a special fixture usually is required for each part.

This process will not solve all problems of warpage in machining. It may reduce warpage internally but increase warpage of the extreme outer layers, although in the opposite direction (Fig. 31). Also, the effect of the altered residual stress pattern on performance must be evaluated carefully for each part. This is particularly important for parts subjected to cyclic loading or exposed to corrosive environments such as marine atmospheres, especially if the process is introduced after the start of production and original performance tests are not repeated. Further disadvantages are the cost and hazard involved in handling liquid nitrogen and live steam.

Effects of Reheating

The precipitation characteristics of aluminum alloys must be considered frequently during evaluation of the effects of reheating on mechanical properties and corrosion resistance. Such evaluations are necessary for determining standard practices for manufacturing operations (such as hot forming and straightening, adhesive bonding, and paint and dry-film lubricant curing) and for evaluating the effects of both short-term and long-term exposure in elevated temperatures in service.

The stage of precipitation that exists in an alloy at the time of reheating plays a significant role in the effects of reheating. Consequently, it is extremely dangerous to reheat material in a solution heat-treated temper without first carefully testing the effects of such reheating. In one such test, 2024-T4 sheet was found to be very susceptible to intergranular corrosion when subjected to a 15 min drying operation at 150 °C (300 °F) during the first 8 h after quenching. No susceptibility was evident when the same drying operation was performed more than 16 h after quenching. In another test, 7075-W (0.2 to 600 h) bar and plate were reheated for hot forming at 175 °C (350 °F) for 20 min. Strengths after aging to the T6 temper were 10 to 15% lower than those for standard 7075-T6. In contrast, similar reheating of T6 material for up to 1 h at 175 °C (350 °F) produced no detrimental effect.

If reheating is performed on material in the W or T4 condition, its effect can be estimated

Table 13 Typical full annealing treatments for some common wrought aluminum alloys
These treatments, which anneal the material to the O temper, are typical for various sizes and methods of manufacture and may not exactly describe optimum treatments for specific items

Alloy	Metal temperature °C	Metal temperature °F	Approximate time at temperature, h
1060	345	650	(a)
1100	345	650	(a)
1350	345	650	(a)
2014	415(b)	775(b)	2–3
2017	415(b)	775(b)	2–3
2024	415(b)	775(b)	2–3
2036	385(b)	725(b)	2–3
2117	415(b)	775(b)	2–3
2124	415(b)	775(b)	2–3
2219	415(b)	775(b)	2–3
3003	415	775	(a)
3004	345	650	(a)
3105	345	650	(a)
5005	345	650	(a)
5050	345	650	(a)
5052	345	650	(a)
5056	345	650	(a)
5083	345	650	(a)
5086	345	650	(a)
5154	345	650	(a)
5182	345	650	(a)
5254	345	650	(a)
5454	345	650	(a)
5456	345	650	(a)
5457	345	650	(a)
5652	345	650	(a)
6005	415(b)	775(b)	2–3
6009	415(b)	775(b)	2–3
6010	415(b)	775(b)	2–3
6053	415(b)	775(b)	2–3
6061	415(b)	775(b)	2–3
6063	415(b)	775(b)	2–3
6066	415(b)	775(b)	2–3
7001	415(c)	775(c)	2–3
7005	345(d)	650(d)	2–3
7049	415(c)	775(c)	2–3
7050	415(c)	775(c)	2–3
7075	415(c)	775(c)	2–3
7079	415(c)	775(c)	2–3
7178	415(c)	775(c)	2–3
7475	415(c)	775(c)	2–3
Brazing sheet			
No. 11 and 12	345	650	(a)
No. 21 and 22	345	650	(a)
No. 23 and 24	345	650	(a)

(a) Time in the furnace need not be longer than necessary to bring all parts of the load to appealing temperature. Cooling rate is unimportant. (b) These treatments are intended to remove the effects of solution treatment and include cooling at a rate of about 30 °C/h (50 °F/h) from the annealing temperature to 260 °C (500 °F). Rate of subsequent cooling is unimportant. Treatment at 345 °C (650 °F), followed by uncontrolled cooling, may be used to remove the effects of cold work or to partly remove the effects of heat treatment. (c) These treatments are intended to remove the effects of solution treatment and include cooling at an uncontrolled rate to 205 °C (400 °F) or less, followed by reheating to 230 °C (450 °F) for 4 h. Treatment at 345 °C (650 °F), followed by uncontrolled cooling, may be used to remove the effects of cold work or to partly remove the effects of heat treatment. (d) Cooling rate to 205 °C (400 °F) or below is less than or equal to 30 °C/h (50 °F/h).

from families of precipitation heat-treating curves such as those presented in Fig. 23. Such curves can also be used for reheating of precipitation heat-treated material at the precipitation heat-treating temperature. For reheating at other temperatures, other data may be needed (Fig. 32). The heat treating and reheating curves may be used as the bases for limitations on reheating (Table 11).

Annealing

Annealing treatments employed for aluminum alloys are of several types that differ in objective. Annealing times and temperatures depend on alloy type as well as on initial structure and temper.

Full Annealing. The softest, most ductile, and most workable condition of both non-heat-treatable and heat-treatable wrought alloys is produced by full annealing to the temper designated "O." Strain-hardened products in this temper normally become recrystallized, but hot-worked products may remain unrecrystallized. In the case of heat-treatable alloys, the solutes are precipitated sufficiently thoroughly to prevent natural age hardening. A higher maximum temperature and additional holding time at the lower temperature generally are employed.

For both heat-treatable and non-heat-treatable aluminum alloys, reduction or elimination of the strengthening effects of cold working is accomplished by heating at a temperature from about 260 to about 440 °C (500 to 825 °F). The rate of softening is strongly temperature-dependent; the time required to soften a given material by a given amount can vary from hours at low temperatures to seconds at high temperatures.

If the purpose of annealing is merely to remove the effects of strain hardening, heating to about 345 °C (650 °F) will usually suffice. If it is necessary to remove the hardening effects of a heat treatment or of cooling from hot working temperatures, a treatment designed to produce a coarse, widely spaced precipitate is employed. This usually consists of soaking at 415 to 440 °C (775 to 825 °F) followed by slow cooling (28 °C/h, or 50 °F/h, max) to about 260 °C (500 °F). The high diffusion rates that exist during soaking and slow cooling permit maximum coalescence of precipitate particles and result in minimum hardness.

As a result of this treatment, only partial precipitation occurs in 7xxx alloys, and a second treatment (soaking at 230 ± 6 °C, or 450 ± 10 °F, for 2 h) is required. When the need arises for small additional improvements in formability, cooling at 28 °C/h (50 °F/h) should be extended to 230 °C (450 °F), and the material should be soaked at 230 °C (450 °F) for 6 h. The effects of eliminating or prolonging the 230 °C (450 °F) second step on the ductility of 7075-O sheet are compared with the standard treatment in Table 12.

In annealing, it is important to ensure that the proper temperature is reached in all portions of the load; therefore, it is common to specify a soaking period of at least 1 h. The maximum annealing temperature is moderately critical; it is advisable not to exceed 415 °C (775 °F) because of oxidation and grain growth. The heating rate can be critical, especially for alloy 3003, which usually requires rapid heating for prevention of grain growth. Relatively slow cooling, in still air or in the furnace, is recommended for all alloys to minimize distortion. Typical annealing conditions used for some alloys in common use are listed in Table 13.

Products that can be heated and cooled very rapidly, such as wire, are annealed by continuous processes that require a total heating and cooling time of only a few seconds. Continuous annealing of coiled sheet is accomplished in a total time of a few minutes. For these extremely rapid operations, maximum temperature may exceed 440 °C (825 °F).

Although material annealed from the precipitation-hardened condition usually has sufficient ductility for most forming operations, this ductility often is slightly lower than that of material that has not been subjected to prior heat treatment (that is, material annealed at the producing source). Therefore, when maximum ductility is required, annealing of a previously heat-treated product is sometimes unsuccessful.

Partial Annealing. Annealing of cold-worked, non-heat-treatable wrought alloys to obtain intermediate mechanical properties (H2-type tempers) is referred to as partial annealing or recovery annealing. Temperatures used are below those that produce extensive recrystallization, and incomplete softening is accomplished by substructural changes in dislocation density and rearrangement into cellular patterns (polygonization). The bendability and formability of an alloy annealed to an H2-type temper generally are significantly higher than in an identical alloy in which an equal strength level is developed by a final cold-working operation (H1-type temper). Treatments to produce H2-type tempers require close control of temperature to achieve uniform and consistent mechanical properties.

Figure 33 shows changes in yield strength as functions of temperature and time for sheet of two non-heat-treatable alloys (1100 and 5052) initially in the highly cold-worked condition (H18 temper). From these curves, it is apparent that by selecting appropriate combinations of time and temperature, mechanical properties intermediate to those of cold-worked and fully annealed material can be obtained. It is also evident that yield strength depends much more strongly on temperature than on time of heating.

Stress-Relief Annealing. For cold-worked wrought alloys, annealing merely to remove the effects of strain hardening is referred to as stress-relief annealing. Such treatments employ temperatures up to about 345 °C (650 °F), or up to 400 ± 8 °C (750 ± 15 °F) for 3003 alloy, and cooling to room temperature. No appreciable holding time is required. Such treatment may result in full recrystallization or simple recovery partial recrystallization. Age hardening may follow stress-relief annealing of heat-treatable alloys; however, a concentration of soluble alloying elements sufficient to cause natural ag-

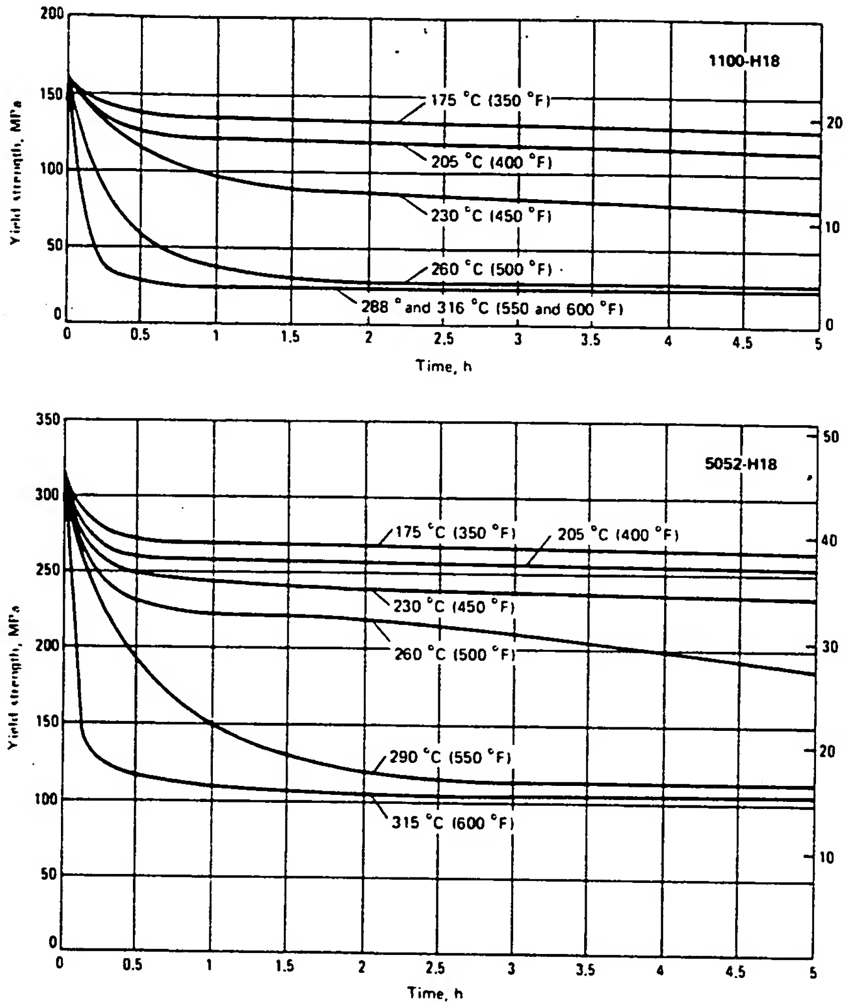

Fig. 33 Representative isothermal annealing curves for alloys 1100-H18 and 5052-H18

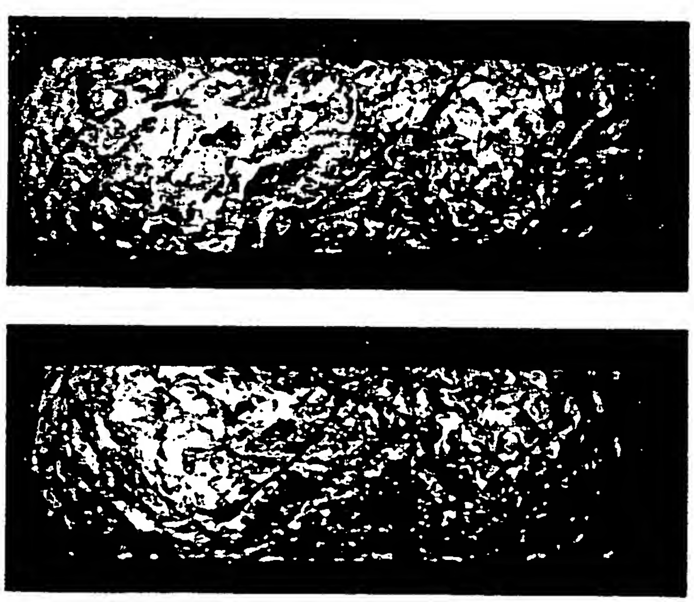

Fig. 34 Two views of a compressed specimen of aluminum alloy 7075-T6 displaying "orange peel" effect. Extensive microcracking is evident in the severely wrinkled surface. Microscopic examination of the surface revealed extensive microcracking in the valleys of the wrinkles.

ing remains in solid solution after such treatments.

A special form of stress-relief temper is used for heat-treatable alloy products that subsequently will be inspected ultrasonically. The product is heated to its normal solution heat-treating temperature, then cooled in still air to room temperature. This temper is referred to as the O1 temper.

Controlled-Atmosphere Annealing and Stabilizing. Aluminum alloys that contain even very small amounts of magnesium will form a surface magnesium oxide unless the atmosphere in the annealing furnace is free of moisture and oxygen. Examples include alloy 3004, which is used for cooling utensils and alloys of the 5*xxx* series.

Another problem that control of the annealing atmosphere helps to overcome or avoid is oil staining by oil-base roll lubricants that do not burn off at lower annealing temperatures. If the oxygen content of the furnace atmosphere is kept very low during such annealing, the oil will not oxidize and stain the work.

Temperature control for full and partial annealing is somewhat more critical than for stress-relief annealing. The temperatures and times specified are selected to produce recrystallization and, in the case of heat-treatable alloys, a precipitate of maximum size; for this the cooling rate must be closely controlled. Even allowing the load to cool in the furnace may result in an excessively high rate. Similarly, lowering the furnace-control instrument by 28 °C (50 °F) each hour may produce stepped cooling, which is not satisfactory for severe forming operations. For maximum softening, a continuous cooling rate of not more than 28 °C/h (50 °F/h) is recommended.

Anealing of castings for 2 to 4 h at temperatures from 315 to 345 °C (600 to 650 °F) provides the most complete relief of residual stresses and precipitation of the phases formed by the excess solute retained in solid solution in the as-cast condition. Such annealing treatments provide maximum dimensional stability for service at elevated temperatures. The annealed temper is designated "O." (This temper was designated "T2" prior to 1975.)

Grain Growth

Many of the aluminum alloys in common use are subject to grain growth during solution treatment or annealing. This phenomenon can occur during or after recrystallization of material that has been subjected to a small critical amount of prior cold work. It is usually manifested by surface roughening during subsequent fabrication operations and frequently results in rejections for appearance or functional reasons. Less frequently, some deterioration of mechanical properties is encountered, and this is undesirable regardless of surface roughening effects.

Degree of susceptibility to grain growth varies with alloy, structure, chemical composi-